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SMALL CRACK GROWTH AND ITS INFLUENCE IN NEAR ALPHA-TITANIUM ALLOYS

by

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SUMMARY

Small fatigue cracks are known to propagate at rates much greater than large cracks in the same material at an equivalent crack driving force. The review clarifies the small crack problem and the reasons for its existence. Some recent models, incorporating large scale plasticity, crack closure and micro-structural dissimilitude, are introduced. Small crack behaviour in near alpha-titanium alloys is examined. Great Britain. (AW)

This Memorandum expresses the opinions of the author and does not necessarily represent the official view of the Royal Aerospace Establishment.

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1 INTRODUCTION

Engine designers are continually searching for higher working temperatures in gas turbine engines in order to improve thrust and to save fuel. The temperatures reached in the engine are limited by the performance of materials. For new materials to be used safely and to their full potential, the initiation and subsequent growth of cracks must be carefully examined so that a fracture mechanics lifting criterion (incorporating defect tolerance) can be adopted. The growth of large cracks in high temperature titanium alloys is relatively well understood under simple loading conditions. However the anomalous behaviour of small cracks needs further investigation.

It is the aim of this review to introduce the small crack effect, compare it with large crack behaviour and examine problems peculiar to titanium alloys.

1.1 Small, short and large cracks

To begin to study the small crack effect it is first necessary to define what is meant by a small crack and differentiate it from short and large cracks in the same material. Over the past decade various conditions have been proposed for observing rapid growth of small fatigue cracks; most of the references quoted in the review have come to more or less the same conclusions but the following, offered by Lankford and Davidson[1] is more specific.

- Crack tip plastic zone r_p is less than the average grain size.
- Length of the crack front is such that it crosses a small number of grains.

Small cracks grow significantly faster than large cracks at equivalent crack driving force, and often below the fatigue threshold ΔK_{th}. Physically short cracks, on the other hand, such as through thickness cracks (less than 2 mm in length) do not grow much faster than large cracks at a given ΔK but do exhibit lower thresholds. As shown in Fig 1 [2] the rate of growth above the threshold tends to follow the large crack curve. Such cracks, not to be mistaken for small cracks, are large with respect to the microstructural unit, usually the grain size.

In comparing small or short cracks with long ones of the same stress intensity, the validity of using linear elastic fracture mechanics (LEFM) and the effects of microstructure and crack closure must be thoroughly examined. In applying LEFM the engineer is assuming that the concept of similitude exists, that is equal changes in the stress intensity factor (crack driving force) produces equal amounts of crack advance. Unfortunately the conditions for

similitude are violated so that ΔK does not fully characterize the crack tip stress fields for small cracks.

2 MECHANICAL MODELLING OF SMALL CRACK BEHAVIOUR

2.1 Similitude

The requirements for similitude are rigorously given elsewhere[3,4] but for simplicity it is sufficient to say that under conditions of equal constraint and equal ΔK , two cracks in the same material should grow at the same rate. For cyclic loading the stress intensity range is commonly related to crack growth rate through power laws of the form,

$$\frac{da}{dN} = \beta \Delta K^c \tag{1}$$

where β and c are experimentally determined constants.

If the stress state is different the constraint is not the same, and so the sizes of the crack tip stress field and plastic zone are not equal. One of the main reasons for the breakdown of similitude lies in the restriction placed on plastic zone size, such that it must be small in comparison to all length dimensions. With this in mind Lankford[5] reported that the small crack effect could occur when the plastic zone size to crack length ratio (r_p/a) exceeded 0.05 but conceded that some small fatigue cracks may grow at the same rate as large ones for r_p as large as 0.24a. The critical factor is the ratio of plastic zone size to the relevent microstructural dimension.

2.2 Microstructural similitude

So far we have only considered the problems invoked by small crack growth with respect to the mechanical aspects of similitude. It is quite clear however that small cracks do not follow the requirements of metallurgical similitude because their crack fronts are short in that they only cover a few grains. Thus the properties of individual grains are important. Large cracks do not suffer from microstructural "dissimilitude" as their crack fronts interrogate a sufficient number of grains so that the fracture properties of differently orientated grains are averaged out.

Similitude in crack growth is not required in comparing the response of small and large cracks although differences do affect the growth rate. Lastly changes in environment do influence the growth mechanism, for instance the

operation of a cleavage-like fracture mode in alpha-titanium is more frequent under a severe environment such as water or sodium chloride solution[6]. The formation of an oxide layer on the fracture surface may affect crack closure.

2.3 Violation of the small scale yielding assumption

The general form of the mode I stress intensity range for a crack under cyclic loading is given by,

$$\Delta K_I = \beta\Delta\sigma(\pi a)^{1/2} \qquad (2)$$

where β is the geometry factor, $\Delta\sigma$ is the the applied stress range, and a is the crack length.

For a smaller crack to have the same ΔK it requires a higher stress, and has a larger plastic zone to crack length ratio. Moreover the stress applied for a small crack to have an equivalent driving force would closely approach the yield stress, making the plastic zone rather large and difficult to define. Once the plastic zone becomes large with respect to the crack size, the small scale yielding assumption breaks down, and so ΔK does not adequately describe the crack tip stress field and the size of the plastic zone.

Presently no single parameter has been established to describe exactly the driving force for small cracks although ΔK has been modified[7], or alternatively determined for specific geometries via finite element and boundary-collocation analysis[8,9]. Elastic plastic fracture mechanics (EPFM)[10,11] and local crack tip field parameters[4,12] have also been introduced.

As the small scale yielding assumption is violated for small cracks, it was thought that the J integral (derived by Rice[13]), which describes the crack tip strain fields, could be a more acceptable parameter. Dowling[10] found that it was possible to correlate small fatigue crack growth using the J integral except for very small surface cracks, less than 0.007 inches ($\simeq$ 178 μm) in length. Its failure to predict the behaviour of these cracks was associated with "non-continuum" effects, as 0.007 inches represents only ten times the average grain size for the material. Tanaka[11] has also considered the cyclic J integral for fatigue crack growth but discovered it to be sensitive to the loading process for large scale yielding.

Similarly Broek[4] has recognised that K is an unsuitable parameter for small crack growth and suggested it be replaced by crack tip strain ($\Delta\varepsilon_p$). The

decision to exploit $\Delta\varepsilon_p$ was made on the basis that a given amount of slip (which is responsible for fatigue crack growth) produces a given amount of crack extension, and equal plastic strain produces equal slip and equal amounts of crack extension. Therefore if there is metallurgical similitude equal plastic strain range should give the same crack growth. An effective ΔK , derived from $\Delta\varepsilon_p$, has been offered to predict the growth of small cracks but conclusive experimental evidence to support this has yet to be presented. However Lankford and Davidson[14] using a technique called stereoimaging analysis have indicated that local crack tip strains for small cracks are larger, and distributed differently, compared to large cracks of equivalent ΔK .

One of the first modifications to ΔK (or rather ΔK_{th}) was in the addition of a plastic zone correction. If ΔK_{th} is taken to be a materials constant then it follows that,

$$\Delta\sigma_{th} = \frac{\Delta K_{th}}{B(\pi a)^{1/2}} \tag{3}$$

where β is a geometric factor.

Examining the equation shows that $\Delta\sigma_{th} \rightarrow \infty$ as $a \rightarrow 0$ but in fact when $a = 0$, $\Delta\sigma_{th} = \Delta\sigma_e$, the endurance limit. On reflection this anomaly can be expected because similitude breaks down for $a = 0$. It was thought that adding a plastic zone correction factor would rectify the problem *ie*

$$\Delta\sigma_{th} = \frac{\Delta K_{th}}{\beta[\pi a + r_p)]^{1/2}} \quad . \tag{4}$$

However when ΔK_{th} is fixed the plastic zone size is constant so that ΔKth is finite when $a = 0$, and obviously the plastic zone size is not r_p for $a = 0$, thus leaving (4) invalid for small cracks. El Haddad *et al*[7] offered a similar expression using an effective crack length $l + l_0$, l_0 being a materials constant. Although some correlation was found between experimental and predicted data, general application of the model is not possible as both β and r_p are sensitive to changes in geometry.

Recent attempts to model small crack growth involve looking more closely at crack tip regions, initially via the assessment of crack tip strains and opening displacements[14,15], found to be larger, and crack opening loads lower, than for large cracks. The fact that crack tip opening displacement (CTOD) for small cracks exceeds those for large cracks at equivalent stress intensities, can partly account for the more rapid growth of small cracks through the relation[16],

$$\frac{da}{dN} \propto \Delta \mathrm{CTOD} \quad . \tag{5}$$

These observations were used to produce local ΔJ and ΔK parameters which were shown to be larger than the applied ΔK for small cracks[12]. A model to predict the local ΔJ and ΔK parameters was also given based on the Barenblatt-Dugdale crack, and compared with experimental data. Agreement between the two sources was not achieved at low applied ΔK, suggesting that other factors such as microstructural effects do contribute to the small crack problem. From a practical viewpoint it is also worth noting that crack tip displacement and strain measurements necessary for evaluating the local parameters are not easily attained.

3 CRACK CLOSURE

It is probably appropiate at this point to deal with crack closure before continuing with more complex models (its effects were largely avoided in the above). Closure can occur even at positive loads during cycling when there is physical contact between mating fracture surfaces in the wake of, or at, the crack tip. The crack will not propagate until K reaches a value K_{op}, at which the crack faces are open. Therefore the effect of closure is to reduce the applied stress intensity range ($K_{max} - K_{min}$) to a lower effective value ΔK_{eff}, given by $K_{max} - K_{op}$. ΔK_{eff} is related to ΔK through the U value, given as $U = \Delta K_{eff}/\Delta K$, which is a function of the stress ratio R ($\sigma_{min}/\sigma_{max}$). Crack closure can be minimized by applying high R values. In support of this McCarver and Ritchie[17] reported that ΔK_{th} values of long and short cracks, measured using $R = 0.8$ were very similar.

The differences in propagation behaviour of long and physically short cracks are almost exclusively due to crack closure. Crystallographic crack growth (*eg* as seen in titanium) leaves faceted fracture surfaces which as a result of combined mode I and mode II displacements, meet at discrete points

wedging open the crack (Fig 2)[17] so that the tip does not close thus preventing the stress intensity falling to the theoretical value calculated from the minimum load. Therefore ΔK is reduced but is greater than the effective value calculated using K_{op}[18]. A short crack on the other hand would experience very little roughness induced crack closure. Hicks *et al*[19] have also shown crack closure in a titanium alloy to be dependent on microstructure.

3.1 Threshold assessment

The large crack threshold (Fig 1) is thought to be an experimental artifact and not a genuine material characteristic. An intrinsic material threshold does exist and its value is sensitive to composition and microstructure[21], but load shedding procedures produce a premature threshold due to closure effects. This was confirmed by Minakawa and McEvily[20] through measurements of crack opening load P_{op} from a compact tension specimen as ΔK approached ΔK_{th}. The ratio of P_{op}/P_{max} increased sharply, almost to unity, near ΔK_{th}. Similarly Lankford and Davidson[1], using data presented by Breat *et al*[2], and James and Knott[21], showed that the K_{op}/K_{max} ratio for large cracks remained approximately constant in the Paris regime but increased as ΔK approached ΔK_{th}. However for short cracks the ratio is essentially constant so that the da/dN versus ΔK relationship appears as an extension of the large crack data below the threshold (Fig 1).

The threshold results from residual plastic deformation in the wake of the crack, formed as a consequence of load shedding. These plastic areas near the surface of the specimen (regions of plane stress) effectively control P_{op} and tend to prop open the crack at the tip (Fig 3). Crack advance will continue once the surfaces are fully apart so that the value of ΔK_{th} obtained will be artificially high. Therefore to study the behaviour of large cracks at low ΔK the load reduction schemes should be altered to remove the residual deformations, perhaps via machining or the application of large compressive loads during threshold tests.

Closure mechanisms are enhanced at near threshold stress intensities without the effects caused by load shedding due to smaller crack tip opening displacements and a greater tendency for cracks to propagate in a crystallographic fashion. The crack will cease to propagate when a given stress intensity is unable to drive it through the material's microstructure.

4 THE ROLE OF MICROSTRUCTURE

For a model of small crack growth to be successful, a third aspect needs to be considered, namely that of microstructure and the effects it has on

similitude. The model will have to explain the differences in behaviour of small cracks, *ie* why some microcracks permanently arrest whilst others retard intermittently or just accelerate with increasing ΔK (Fig 4).

Both of the factors deciding whether the small crack effect will take place or not deal with comparing grain size to the dimensions of the crack or its plastic zone. It is therefore worth considering the response to the two extremes of grain size, that of a very fine microstructure and the single crystal (infinite grain size). Hicks and Brown[22] have studied the behaviour of large and small cracks in fine grained (5 μm), coarse grained (50 μm) and single crystal Astroloy. Their results reproduced in (1) show that crack growth rates in the single crystal are greater than those obtained from the polycrystalline material for a given ΔK (Fig 5). The data for small and large cracks in the single crystal are almost superimposed apart from the threshold feature of the large crack curve caused by using load shedding as a means of growing cracks at low ΔK. This superposition of data is to be expected as the single crystal is capable of accommodating the plastic zone of very large cracks. Incidently, confirmation of the conditions speculated for small crack growth are also produced within the article via the calculated plastic zone size at convergence of small and large crack data for the polycrystalline alloy.

In contrast to the above, small cracks in a very fine grained (7 μm in the direction of crack extension), powder metallurgy aluminium alloy appear to show the same da/dN versus ΔK relationship for large cracks (Fig 6)[1]. The fact that the size of the plastic zone of any microcrack is likely to very quickly exceed the grain size would explain the behaviour.

4.1 Retardation and arrest features of small crack growth

As a consequence of examining crack growth in single crystals it is thought that the unduly "slow" growth in polycrystalline materials is linked to the crack approaching or crossing a grain boundary or phase interface. Fig 7 [7] clearly shows the response of a crack approaching and leaving a grain boundary. After a rise in growth rate towards the boundary the crack arrests for approximately 2000 cycles and then breaks away into the next grain. Morris *et al*[23] suggest the arrest or retardation is caused by two effects, firstly, the difficulty in propagating slip across an interface (which will be exaggerated in planar slip materials), and secondly, the increase in P_{op} as the crack tip crosses the interface. Other authors have thought that the effects are due to slip blockage by grain boundaries[24], or the reduction of $\Delta\varepsilon_p$ at the boundaries of less favourably orientated grains[25].

Wagner *et al*[26] have developed an interesting model for small crack propagation which offers another explanation for crack arrest or retardation. They considered the expansion, outwards, of a surface crack in an alpha-titanium alloy and concluded that when the crack tips encounter a grain boundary and are arrested, the crack, instead, grows further into the interior of the grain (Fig 8). Interior crack growth subsequently stops on meeting a boundary. Meanwhile the surface part of the crack may well be in a position to overcome the boundary blocking its growth. Crack growth is then limited to the surface until another boundary is reached when interior growth will occur. This alternating crack propagation mechanism breaks down when the surface crack length becomes quite large and a semicircular profile is adopted with no significant anomalies in growth rate.

Retardation, however, does not take place on every encounter with a grain boundary. Growth is impeded if there is significant misorientation between the cracked and new grain, *ie* if it is hard to get the crack to grow in the next grain. The cracking mechanism or growth mode may change because of the orientation of the following grain, altering da/dN. This implies that rapid growth is, at least in part, the product of a crack propagating in a preferentially orientated grain. A crack extends via a mechanism involving slip and so within a grain crack growth will proceed when the critical resolved shear stress on the optimum slip system is attained.

4.2 Microstructural dissimilitude - a driving force for small crack growth

As the properties of individual grains need to be examined to assess the small crack effect, microstructural similitude is not maintained. Differences in grain orientation and plastic constraint will lead to a variation with crack length/depth of properties such as yield strength, fracture toughness, and crack growth resistance. Chan and Lankford[27] demonstrated the powerful influence of microstructural "dissimilitude" on small crack growth by modelling the variation of yield strength (σ_{ym}), of grains located within the crack tip plastic zone, with crack size (Fig 9).

Morris *et al*[23,28] have shown that cyclic loading (fatigue) causes a reduction in yield strength of individual surface grains, particularly large grains (Fig 10) although its occurrence also depends on grain shape and size distribution. The surface deformation is inhomogeneous, and is constrained by surrounding elastic grains and 'hard' grain boundaries. The authors, in assessing the effect of reduced σ_y in individual grains on crack opening displacement (δ), have presented an expression for δ as a function of grain size and

crack length. Consequently they predict that, for the aluminium alloy studied, δ will increase as σ_y decreases with increasing grain size, above a grain size of 60 µm. However their equation was derived for specific conditions only and hence its application is limited. Softening of certain surface grains may, at least in part, explain the fast growth of small surface fatigue cracks but it is unclear whether the increase in δ is responsible.

5 A POTENTIAL MODEL FOR SMALL CRACK GROWTH

In previous sections of the review the factors which need to be considered for an accurate prediction of small crack behaviour have been identified *ie*

- Large scale plasticity,
- crack closure and,
- microstructural dissimilitude.

A recent attempt to incorporate these points into a model has been made by Chan and Lankford[27]. They have used the Dugdale-Barenblatt crack representation, illustrated in Fig 11, where a small crack is replaced by a larger crack so that a compressive (cohesive) stress σ_{ym} acts over the original process zone. Under these circumstances the crack violates microstructural similitude and the small scale yielding criteria. However the crack can be represented by two loading conditions (Fig 12): (i) a compressive yield stress σ_y, for small scale yielding, and (ii) a tensile stress σ_{dm}, given by $\sigma_{dm} = \sigma_y - \sigma_{ym}$. Thus a small crack which lacks similitude can be viewed as one that obeys LEFM conditions but is subjected to an additional surface traction included as a result of microstructural dissimilitude. The local $\Delta K(\Delta K^*)$ is given by,

$$\Delta K^* = \Delta K + \Delta K_d \qquad (6)$$

where ΔK is the applied stress intensity factor and ΔK_d is the local stress intensity induced as a result of microstructural dissimilitude. The authors then proceeded to evaluate ΔK_d to produce an expression for ΔK^*,

$$\Delta K^* = \Delta K_{eff}\, q_1\, q_2 \qquad (7)$$

where q_1 and q_2 are functions that refer to large scale plasticity (derived by Chan[12]) and microstructural dissimilitude respectively. ΔK_{eff}, of course, accounts for crack closure.

Experimental data from an aluminium alloy were used to support the model and showed it could predict, qualitatively, the rapid extension, retardation and arrest features of small cracks.

6 CRACK GROWTH MECHANISMS PECULIAR TO NEAR ALPHA-TITANIUM ALLOYS

Beta heat treated near alpha-titanium alloys, such as IMI 685 and 829, show very good creep resistance at the service temperatures of aero-engine compressor discs. The microstructure of these alloys is determined by the cooling rate following solution treatment above the beta transus. The martensite structure offers the best combination of creep resistance, tensile strength and ductility but requires, for its formation, cooling rates which exceed those practically possible for large engine components. Instead the basketweave or alpha Widmanstatten structure, is used although colonies of aligned alpha plates of the same crystallographic orientation can be found in some sections experiencing lower cooling rates. The rationale for the growth of these morphologies and the responsible heat treatments are given elsewhere for IMI 685 [29].

Low cycle fatigue and small crack growth studies for IMI 829 are presently in progress within the aero-engine manufacturers and supporting bodies, so the review must concentrate on the established fatigue behaviour of the earlier IMI 685 alloy, although the amount of small crack work, even on this alloy, is limited to a handful of papers[22,26,30-34]. Fortunately large fatigue crack propagation behaviour and LCF is well documented[6,35-37].

Fully heat treated components in IMI 685 are likely to contain some grains which will approach 1 to 2 millimetres in size, so the small crack effect may be observed up to relatively large crack sizes. Taylor and Knott[38] predict that anomalously fast growth will occur below a crack length l_2, for which an estimation of ten times the microstructural unit size (eg grain size), is given. Applying this empirical relation shows that the small crack effect could continue up to crack lengths of a centimetre or more. Therefore the majority of crack propagation prior to fast fracture may be spent in the small crack regime.

The few investigations of small crack growth in IMI 685 indicate that growth rates are greater than those reported in other materials[31,32]. The coarse microstructure is responsible because it presents fewer grain boundaries to retard crack propagation. However the high proportion of high angle boundaries, and the planar nature of slip inherent in the material, will tend to impede

growth more effectively. Cracks were frequently found to arrest at colony boundaries before propagating in a favourable direction in an adjacent colony.

Brown and Hicks[30] have demonstrated that growth rate is sensitive to growth mode, such that increasing da/dN is observed from non-crystallograghic, colony boundary separation and crystallographic growth. Fig 13 shows that crystallographic growth (which results from cracking of intense slip bands on or near the basal plane of the aligned alpha plates) produces higher growth rates than the colony boundary separation propagation mode. Therefore crack growth rate is dependent upon the orientation of material presented to the crack tip.

Much of the differences in behaviour of small and large cracks in near alpha-titanium alloys can be attributed to crack closure. Roughness induced closure is particularly evident from the faceted fracture surfaces. However as suggested earlier, the use of ΔK_{eff} may not correlate small and large crack data in titanium or its alloys.

7 CONCLUSIONS

Small crack growth is a complex and multi-faceted problem. Models that aim to explain small crack behaviour must include the factors mentioned in the review and stay within the laws of similitude. When cracks are microstructurally small the properties of individual grains and the random orientation of neighbouring grains is important. Ultimately a statistical approach may well be required to simulate the propagation of a small crack through microstructure.

Future work on near alpha-titanium alloys is needed to determine the point where small and large crack data converge, and to address the effects of elevated temperature, dwell at maximum load, and aggressive environments so that the fatigue life of gas turbine components can be more accurately predicted.

Appendix

(A) Alternative methods of presenting small/short crack data

Many authors have seen it fit to present small and short crack data in terms of the stress range for fatigue failure versus crack length (Fig 14). The first diagrams, produced by Kitagawa and Takahashi[39], consisted of two lines. The first is for constant stress, so that at zero crack length the stress range is equal to the endurance limit ($\Delta\sigma_e$). The second, given by ΔK_{th}, is for constant stress intensity and is independent of crack length. If LEFM assumptions apply a crack under the influence of a stress range below the ΔK_{th} line should not grow. Note that the small scale yielding criteria tends to be violated above 2/3 of the cyclic yield strength (σ_{cy}) hence the use of elastic plastic fracture mechanics (EPFM) is recommended.

The point where the two lines intersect defines the effective crack length parameter l_0 offered by El Haddad[7]. Experimental short crack data in fact falls between the two lines and defines the lengths l_1 and l_2. Taylor and Knott[38] suggest that a crack of length between l_1 and l_2 shall grow faster, and from a lower ΔK_{th}, than a crack > l_2. They also indicate that small and large crack data converge at l_2.

Miller[40] has shown three regimes of short crack growth on a diagram based on the above (Fig 15). The regimes represent,

(i) small cracks - which propagate below $\Delta\sigma_e$ and are small compared to microstructural dimensions (lines d_1, d_2, d_3),

(ii) physically short cracks - small scale yielding assumption may be invalid,

(iii) highly stressed cracks - length greater than l. EPFM required.

Note that a crack propagating from a notch root is also considered.

From the basic format of the diagram, Brown[41] has constructed fatigue fracture mode maps, with growth rate contours and fatigue fracture modes, for particular materials of specified grain size (Fig 16). If the accuracy of the maps can be guaranteed then they could prove very useful in predicting the behaviour of a given length of crack, under a particular stress range. However as we do not understand all the aspects of small crack growth, I doubt the reliability of such maps and their application to less than very specific circumstances.

(B) The measurement of useful parameters - a review of experimental techniques

Crack tip parameter data, useful in analysing the propagation of small cracks, is not readily available due to the degree of difficulty experienced in their measurement. Table 1 briefly describes some of the techniques discovered in conducting this literature review. Note familiar crack monitoring methods such as the dc potential drop technique (crack length) and crack profile replication techniques have been omitted.

Table 1

SOME UNFAMILIAR EXPERIMENTAL TECHNIQUES

Parameter	Technique	Reference
CTOD	(i) using SEM micrographs	15,23,42
CMOD	(i) using ac LVDT transducer	18,19
	(ii) using clip extensometer	2
	(iii) using interferometric displacement gauge	43
P_{op}	(i) from load v CMOD plots	2,43
	(ii) from stereoimaging photographs	14
	(iii) from Morris equation	42
R_p	(i) from electron channeling patterns obtained from SEM	15
Crack tip strain	(i) using stereoimaging analysis	14

CTOD: crack tip opening displacement

CMOD: crack mouth opening displacement

P_{op}: crack opening or crack closing load

R_p: plastic zone size

REFERENCES

No.	Author	Title etc
1	J. Lankford D.L. Davidson	The Role of Metallurgical Factors in Controlling the Growth of Small Fatigue Cracks. In: Small Fatigue Cracks. Proceedings of the Second Int Conf, (Ed by R.O. Ritchie and J. Lankford) The Metallurgical Society/AIME, Warrendale, Pa, USA, pp 51-71 (1986)
2	J.L. Breat F. Mudry A. Pinneau	Short Crack Propagation and Closure Effects in A508 Steel. Fatigue of Engineering Materials and Structures, **6**, pp 349-358 (1983)
3	B.N. Leis	Discussion of the Short Crack Effect in Airframe Materials and Components. In: Behaviour of Short Cracks in Airframe Components, Proceedings of the Fifty-fifth Meeting of the AGARD Structures and Components Panel, Toronto, Canada, CP-328 (1982)
4	D. Broek	A Similitude Criterion for Fatigue Crack Growth Modelling. In: Fracture mechanics, Sixteenth Symposium, ASTM STP 868, (Ed M.F. Kanninen and A.T. Hopper), American Society of Testing and Materials, Philadelphia, USA, pp 347-360 (1985)
5	J. Lankford	The Influence of Microstructure on the Growth of Small Fatigue Cracks. Fatigue Fract Engng Mater Struct, **8**, pp 161-175 (1985)
6	C.M. Ward-Close C.J. Beevers	The Influence of Grain Orientation on the Mode and Rate of Fatigue Crack Growth in Alpha-Titanium. Met Trans, **11**, pp 1007-1017 (1980)
7	M.H. El Haddad K.N. Smith T.H. Topper	Fatigue Crack Propagation of Short Cracks. ASME, **101**, pp 42-46 (1979)

REFERENCES (continued)

No.	Author	Title etc
8	A.C. Pickard	Stress Intensity for Cracks With Circular and Elliptical Crack Fronts. In: Second Int Conf on Numerical Methods in Fracture mechanics, Swansea, UK, pp 599-614 (1980)
9	J.C. Newman Jr	A Nonlinear Fracture Mechanics Approach to the Growth of Small cracks. In: Behaviour of Short Cracks in Airframe Components, Proceedings of the Fifty-fifth Meeting of the AGARD Structures and Components Panel, Toronto, Canada, CP-328 (1982)
10	N.E. Dowling	Crack Growth During Low Cycle Fatigue of Smooth Axial Specimens. In: Cyclic Stress-Strain and Plastic Deformation Aspects of Fatigue Crack Growth, ASTM STP 637, American Society of Testing and Materials, pp 97-121 (1977)
11	K. Tanaka	The Cyclic J-Integral as a Criterion for Fatigue Crack Growth. Int Journ of Fracture, **22**, pp 91-104 (1983)
12	K.S. Chan	Local Crack-Tip Field Parameters for Large and Small Cracks: Theory and Experiment. In: Small Fatigue Cracks, Proceedings of the Second Int Conf, (Ed by R.O. Ritchie and J. Lankford) The Metallurgical Society/AIME, Warrendale, Pa, USA, pp 407-425 (1986)
13	J.R. Rice	A Path Independent Integral and the Approximate Analysis of Strain Concentration by Notches and Cracks Journal of Applied Mechanics, Transactions of the American Society of Mechanical Engineers, **35**, pp 379-386 (1968)
14	J. Lankford D.L. Davidson	Near-Threshold Crack Tip Strain and Crack Opening for Large and Small Fatigue Cracks. In: Fatigue Crack Threshold Concepts, (Ed by D.L. Davidson and S. Suresh) The Metallurgical Society/AIME, Warrendale, Pa, USA, pp 447-463 (1984)

REFERENCES (continued)

No.	Author	Title etc
15	J. Lankford D.L. Davidson K.S. Chan	The Influence of Crack Tip Plasticity in Growth of Small Fatigue Cracks. Metall Trans, **15A**, pp 1579-1588 (1984)
16	R.O. Ritchie S. Suresh	Mechanics and Physics of the Growth of Small Cracks In: Behaviour of Short Cracks in Airframe Components, Proceedings of the Fifty-fifth Meeting of the AGARD Structures and Components Panel, Toronto, Canada, CP-328 (1982)
17	J.F. McCarver R.O. Ritchie	Fatigue Crack Propagation Thresholds for Long and Short Cracks in Rene 95 Nickel-Base Superalloy, Mat Sci and Eng, **55**, pp 63-67 (1982)
18	N. Walker C.J. Beevers	A Fatigue Crack Closure Mechanism in Titanium Fatigue of Engineering Materials and Structures, **1**, pp 135-148 (1979)
19	M.A. Hicks R.H. Jeal C.J. Beevers	Slow Fatigue Crack Growth and Threshold Behaviour in IMI 685. Fatigue of Engineering Materials and Structures, **6**, pp 51-65 (1983)
20	K. Minakawa A.J. McEvily	On Near-Threshold Fatigue Crack Growth in Steels and Aluminium Alloys. In: Proceedings of the Int Conf on Fatigue Thresholds, Stockholm, Sweden, **2**, pp 36 (1981)
21	M.N. James J.F. Knott	An Assessment of Crack Closure and the Extent of the Short Crack Regime in Q1N (HY80) Steel. Fatigue Fract Engng Mater Struct, **8**, pp 177-182 (1985)
22	M.A. Hicks C.W. Brown	A Comparison of Short Crack Growth Behaviour in Engineering Alloys. In: Proceedings of the Second Int Conf on Fatigue and Fatigue Thresholds, (Ed by C.J. Beevers *et al*), Birmingham, England, **3**, pp 1337-1347 (1984)
23	W.L. Morris M.R. James O. Buck	Growth Rate Models for Short Surface Cracks in Al 2219-T851. Met Trans, **15A**, pp 57-64 (1981)

REFERENCES (continued)

No.	Author	Title etc
24	K.J. Tanaka	Modelling of Propagation and Non-propagation of Small Fatigue Cracks. In: Small Fatigue Cracks, Proceedings of the Second Int Conf, (Ed by R.O. Ritchie and J. Lankford) The Metallurgical Society/AIME, Warrendale, Pa, USA, pp 343-361 (1986)
25	K.S. Chan J. Lankford	A Crack Tip Strain Model for the Growth of Small Fatigue Cracks. Scripta Metall, 17, pp 529-532 (1983)
26	L. Wagner J.F. Gregory A. Gysler G. Lutjering	Propagation Behaviour of Short Cracks in a Ti-8.6Al Alloy. In: Small Fatigue Cracks, Proceedings of the Second Int Conf, (Ed by R.O. Ritchie and J. Lankford) The Metallurgical Society/AIME, Warrendale, Pa, USA, pp 117-128 (1986)
27	K.S. Chan J. Lankford	The Role of Microstructural Dissimilitude in Fatigue and Fracture of Small Cracks. Acta metall, **36**, pp 193-206 (1988)
28	M.R. James W.L. Morris	The Effect of Microplastic Surface Deformation on the Growth of Small Cracks. In: Small Fatigue Cracks, Proceedings of the Second Int Conf, (Ed by R.O. Ritchie and J. Lankford) The Metallurgical Society/AIME, Warrendale, Pa, USA, pp 145-156 (1986)
29	P.J. Postans R.H. Jeal	Dependence of Crack Growth Performance Upon Structure in Beta Processed Titanium Alloys. In: Forging and Properties of Aerospace Materials, Chameleon Press, London, pp 192-198 (1978)
30	C.W. Brown M.A. Hicks	A Study of Short Fatigue Crack Growth Behaviour in Titanium Alloy IMI 685. Fatigue of Engineering Materials and Structures, **6**, pp 67-76 (1983)
31	M.A. Hicks C.W. Brown	Short Fatigue Crack Growth in Planar Slip Materials. Int J Fatigue, **4**, pp 167-169 (1982)

REFERENCES (continued)

No.	Author	Title etc
32	C. Howland M.A. Hicks R.H. Jeal	The Relevance of Short Crack Behaviour to the Integrity of Major Rotating Aero-Engine Components. In: Small Fatigue Cracks, Proceedings of the Second Int Conf, (Ed by R.O. Ritchie and J. Lankford) The Metallurgical Society/AIME, Warrendale, Pa, USA, pp 607-622 (1986)
33	C.W. Brown J.E. King	The Relevance of Microstructural Influences in the Short Crack Regime to Overall Fatigue Resistance In: Small Fatigue Cracks, Proceedings of the Second Int Conf, (Ed by R.O. Ritchie and J. Lankford) The Metallurgical Society/AIME, Warrendale, Pa, USA, pp 73-95 (1986)
34	P.J. Hastings M.A. Hicks J.E. King	The Effect of Alpha Platelet Morphology and Beta Grain Size on the Initiation and Growth of Short Fatigue Cracks in Ti 65S. In: Fatigue '87, Proceedings of Third Int Conf on Fatigue and Fatigue Threshold **I**, (Ed by R.O. Ritchie and E.A. Starke, Jr), EMAS, Warley, England, pp 251-259 (1987)
35	D. Eylon J.A. Hall	Fatigue Behaviour of Beta Processed Titanium Alloy IMI 685. Met Trans, **8A**, pp 981-990 (1977)
36	G.R. Yoder L.A. Cooley T.W. Crooker	Observations on Microstructurally Sensitive Fatigue Crack Growth in a Widmanstatten Ti-6Al-4V Alloy. Met Trans, **8A**, pp 1737-1743 (1977)
37	C.C. Wojcik K.S. Chan D.A. Koss	Stage 1 Fatigue Crack Propagation in a Titanium Alloy. Acta Metall, **36**, pp 1261-1270 (1988)
38	D. Taylor J.F. Knott	Fatigue Crack Propagation Behaviour of Short Cracks; the Effect of Microstructure. Fatigue of Engineering Materials and Structures, **4**, pp 147-155 (1981)

REFERENCES (concluded)

No.	Author	Title etc
39	H. Kitagawa S. Takahashi	Applicability of Fracture Mechanics to Very Small Cracks or the Cracks in the Early Stage. In: Proceedings of Second Int Conf Mech Beh of Mater (ICM2), Boston, USA, American Society of Metals, pp 627-631 (1976)
40	K.J. Miller	The Behaviour of Short Fatigue Cracks and Their Initiation - Part ii - A General Summary. Fatigue Fract Engng Mater Struct, **10**, pp 93-113 (1987)
41	M.W. Brown	Interfaces Between Short, Long and Non-propagating Cracks. Ibid, pp 423-439.(1986)
42	W.L. Morris	The Noncontinuum Crack Tip Deformation Behaviour of Surface Microcracks. Met Trans, **11A**, pp 1117-1123 (1980)
43	J.M. Larsen J.R. Jira T. Weerasooriya	Crack Opening Displacement Measurements on Small Cracks in Fatigue. In: Fracture Mechanics, Eighteenth Symposium, (Ed by D.T. Read and R.P. Reed), ASTM STP 945, American Society for Testing and Materials, Philadelphia, USA, pp 896-912 (1988)

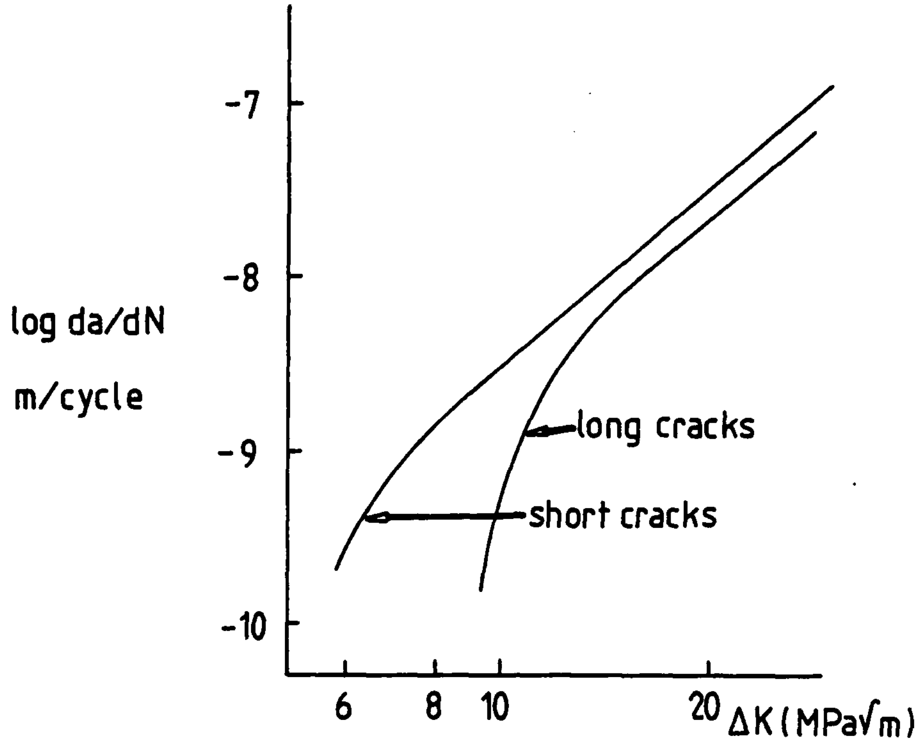

Fig 1 Propagation behaviour of short (0.3 to 0.5 mm) and long cracks (13 to 16 mm) cracks in A508 steel[2]

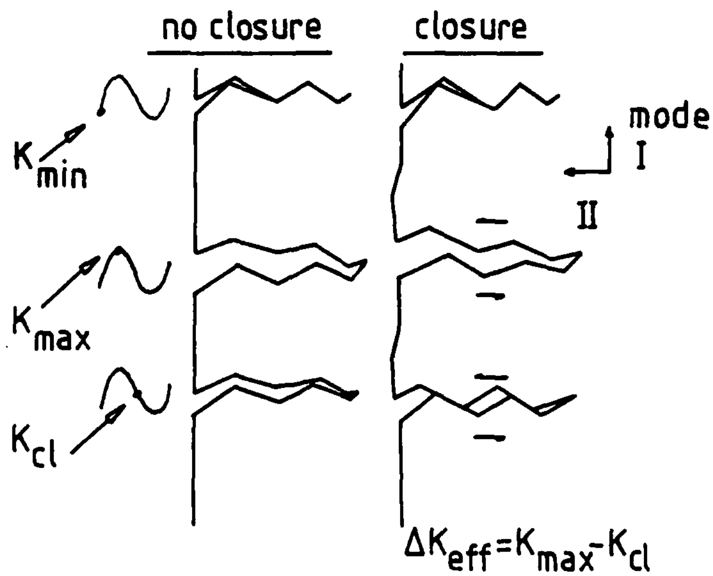

Fig 2 Illustration of roughness - induced crack closure due to combined mode I and mode II crack tip displacements[17]

Figs 3&4

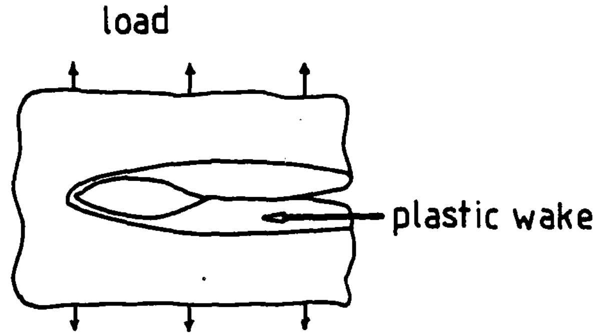

Fig 3 Diagram showing areas of residual plastic deformation responsible for crack closure behind the tip

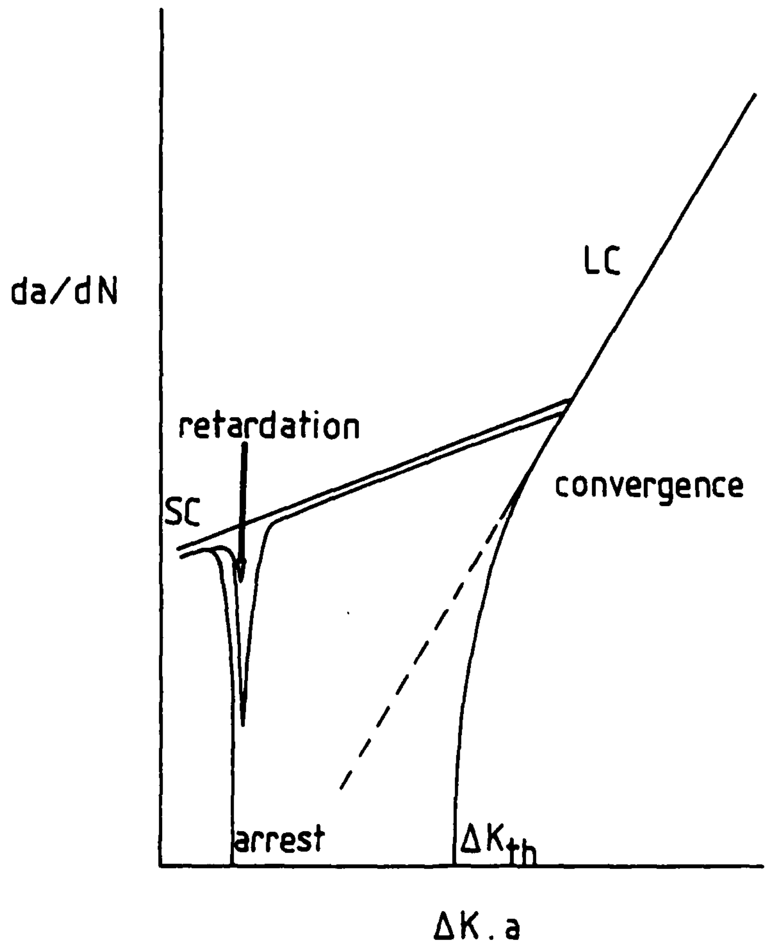

Fig 4 Schematic showing the retardation and arrest features common to small cracks[1]

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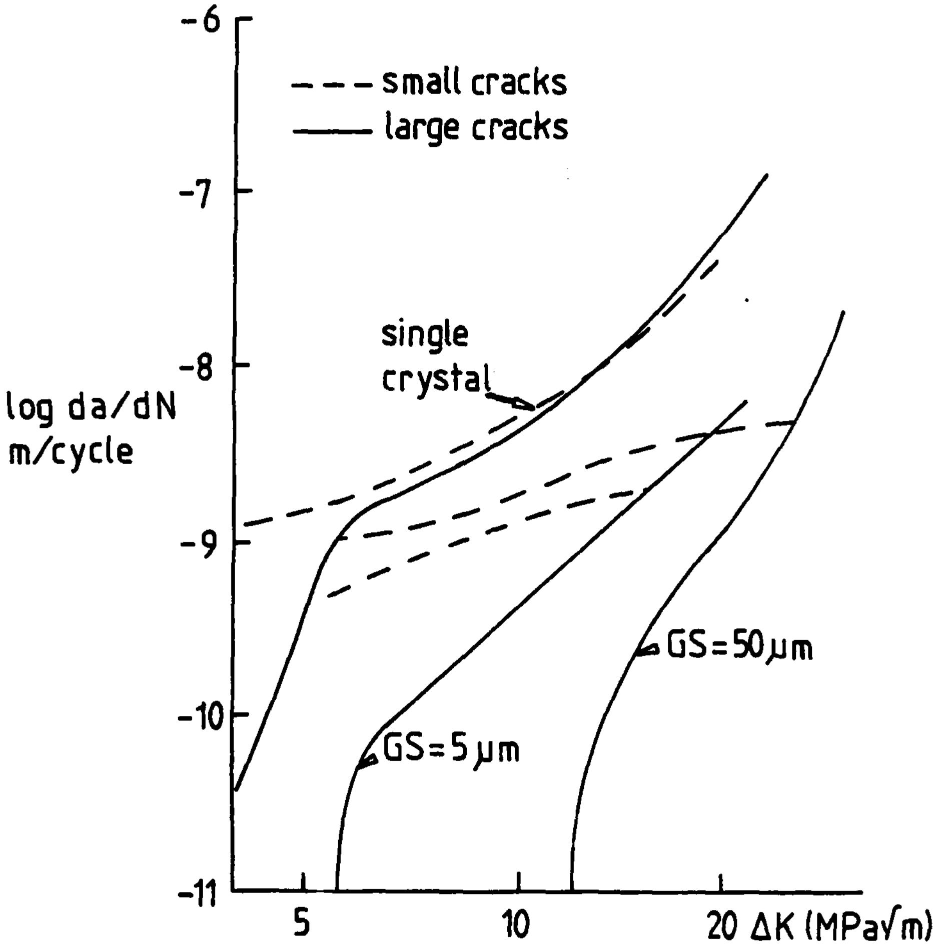

Fig 5 Crack growth rates for small and large cracks in a fine grained (5 μm), coarse grained (50 μm), and single crystal astroloy[22,1]

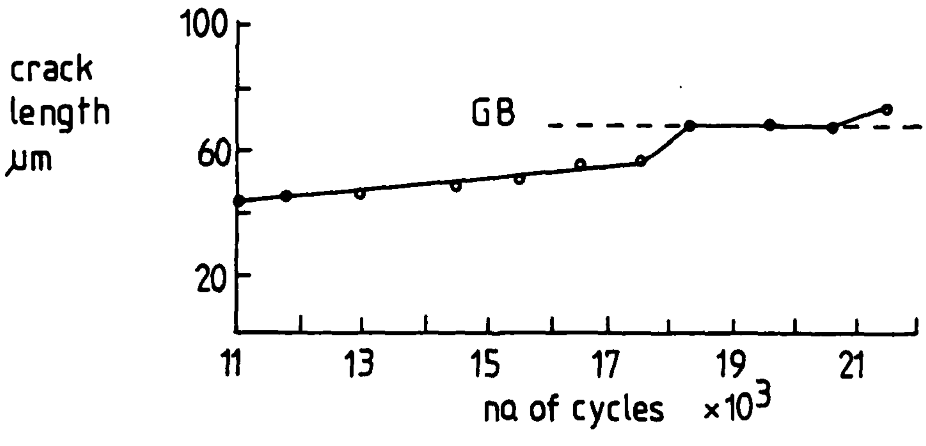

Fig 7 Crack length versus number of cycles showing the arrest of a small crack at a grain boundary[5]

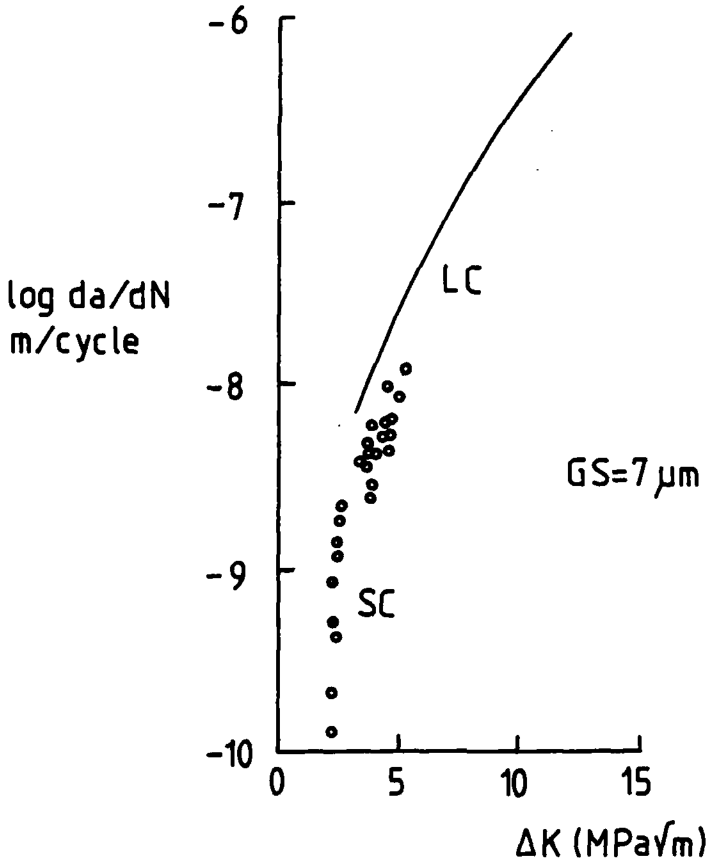

Fig 6 Growth of large and small cracks in P/M produced Al alloy[1]

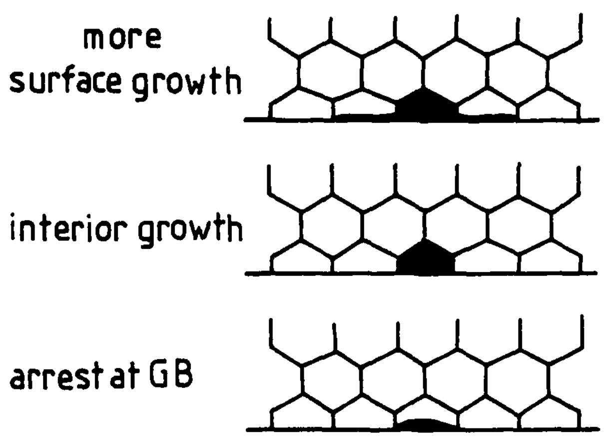

Fig 8 Schematic of the crack propagation mechanism given by Wagner *et al*[26]

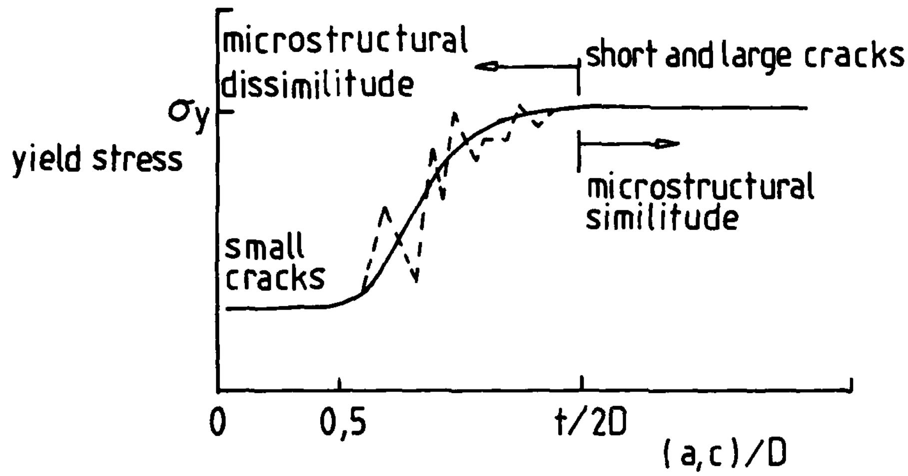

Fig 9 Schematic showing the dependence of yield stress within the plastic zone on the a/D ratio for small, short, and large cracks[27]

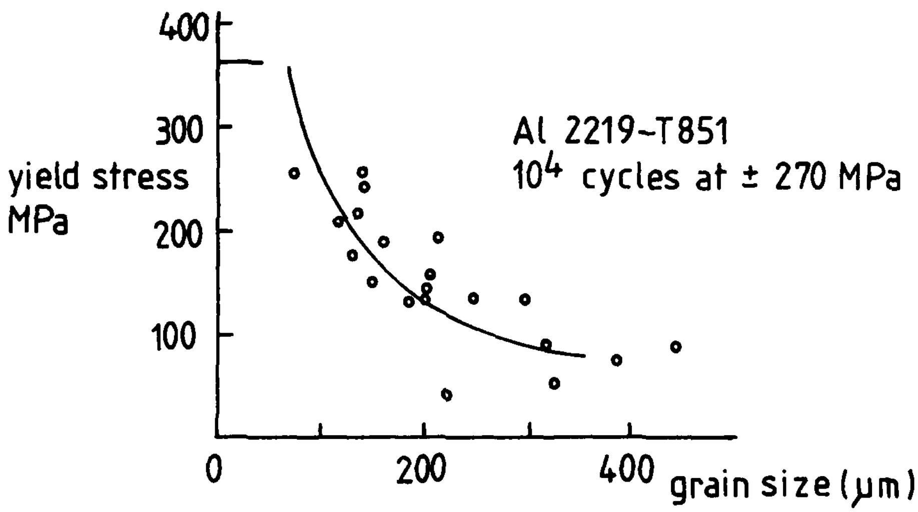

Fig 10 Graph of flow stress versus size of surface grains for Al alloy[23]

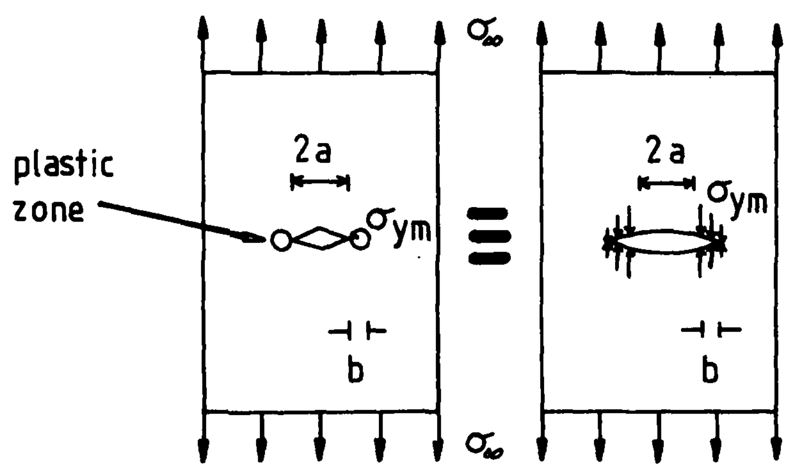

Fig 11 Dugdale-Barenblatt representation of a large crack under small scale yielding conditions[27]

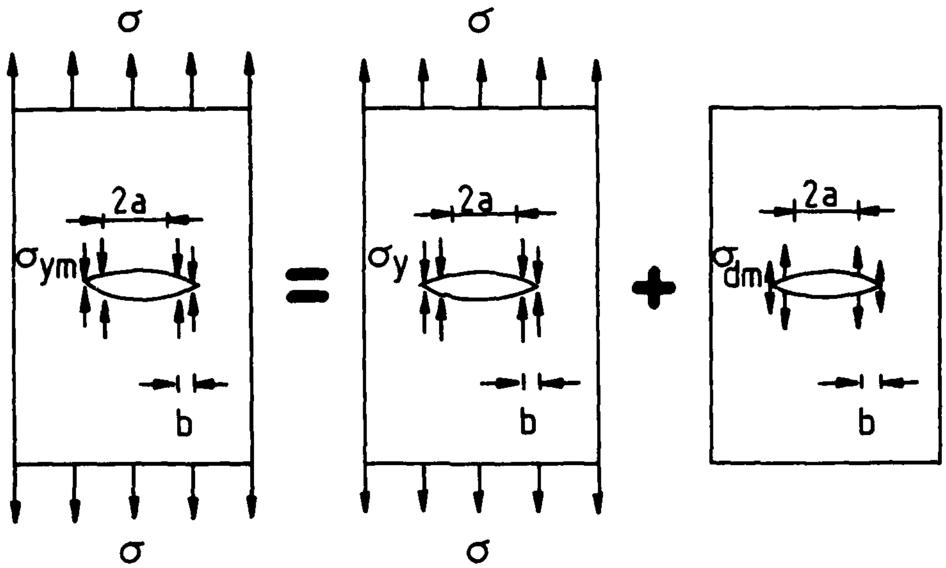

Fig 12 Dugdale-Barenblatt crack represented by 2 loading conditions:
(i) a compressive yield stress σ_y, for small scale yielding, and
(ii) a tensile σ_{dm}, given by $\sigma_{dm} = \sigma_y - \sigma_{ym}$[27]

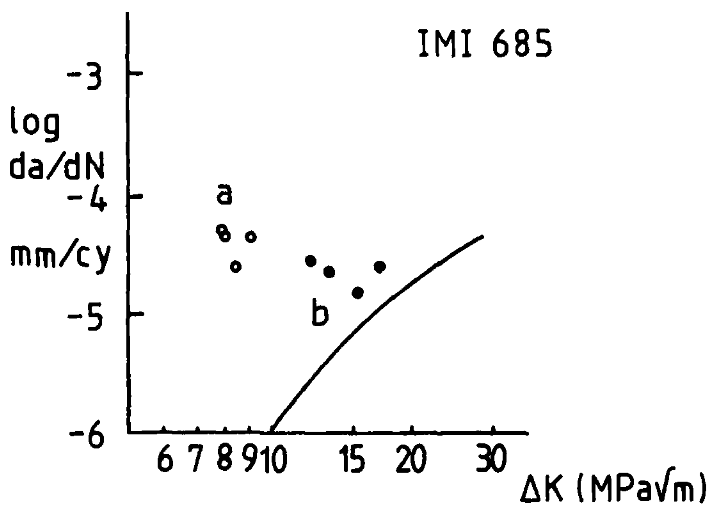

Fig 13 Microstructure and crack growth behaviour of β-processed near alpha-titanium alloy. The figures demonstrate that growth rate is sensitive to growth mode (crystallographic, (a) colony boundary separation (b) or non-crystallographic)[30]

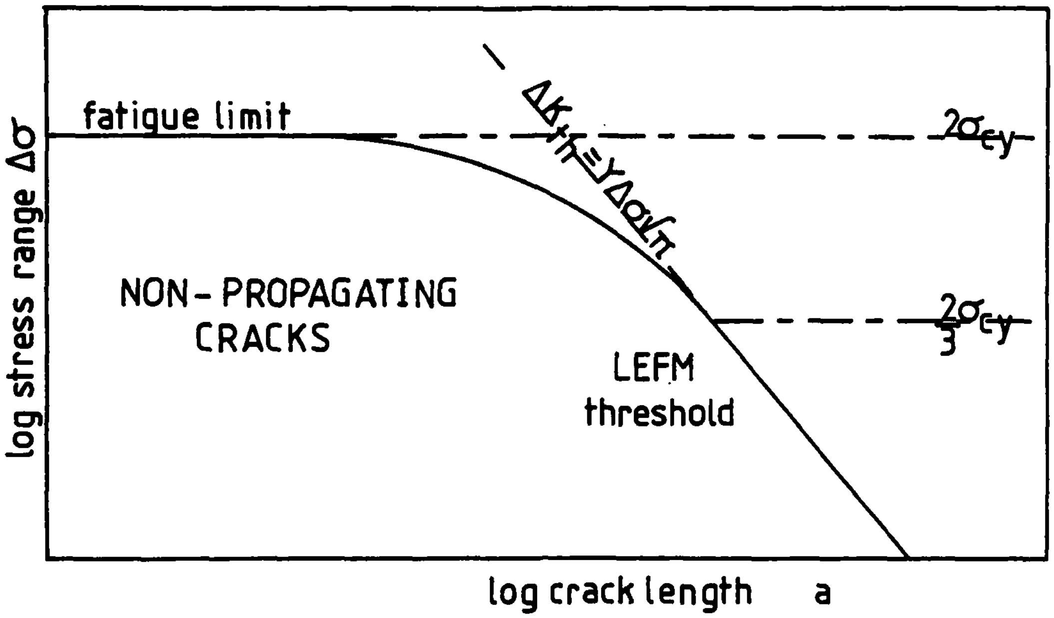

Fig 14 Diagram of stress range versus crack length, with lines for constant stress and ΔK. The curve was established from data given by Kitagawa and Takahashi[40]

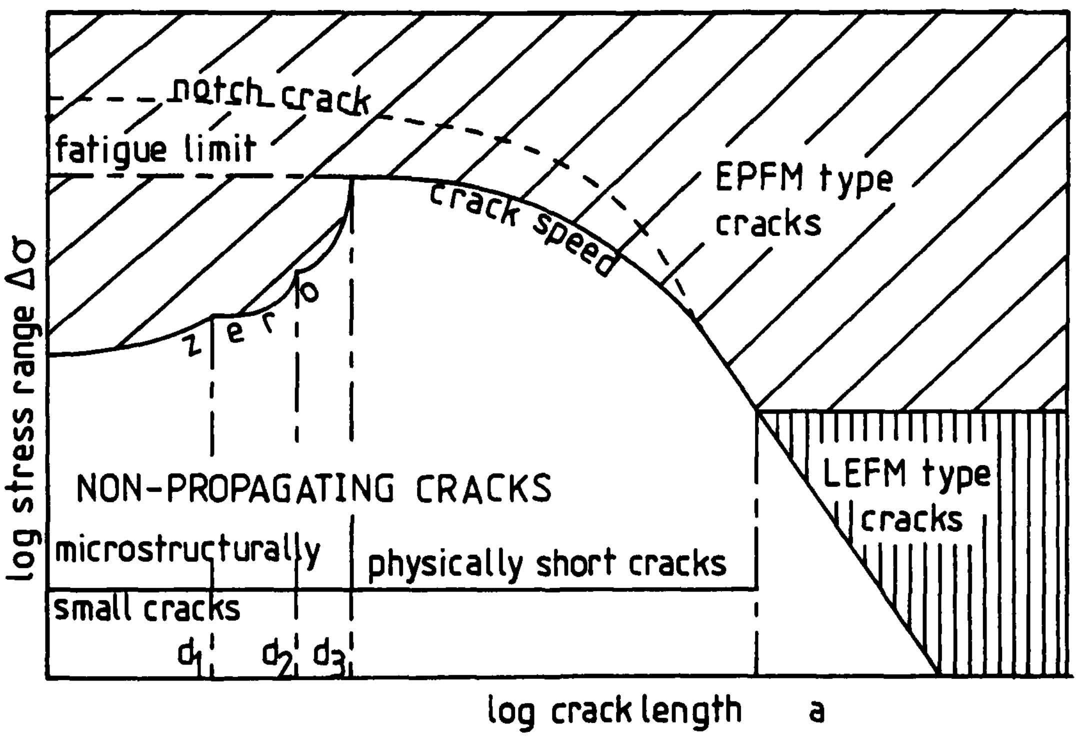

Fig 15 A modified $\Delta\sigma$ versus a diagram showing the regimes of small and short cracks[40]

Fig 16

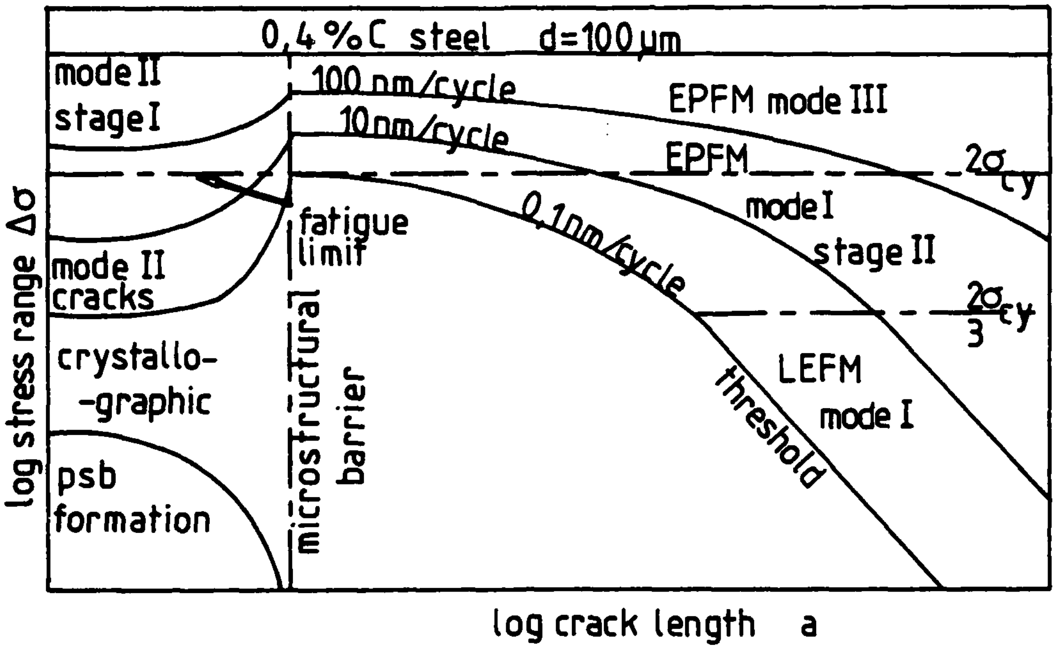

Fig 16 A Brown fatigue fracture mode map for 0.4% C Steel[41]